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ULTRASONIC CHARACTERIZATION OF ALUMINUM MATRIX COMPOSITE ELASTICITY: EXPERIMENT AND THEORY

BY G. V. BLESSING W. L. ELBAN J. V. FOLTZ

RESEARCH AND TECHNOLOGY DEPARTMENT

1 NOVEMBER 1978

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The ability to make meaningful elastic moduli measurements ultrasonically is reported for potential application to the nondestructive evaluation of aluminum matrix composites. The materials were monitored for a change in elasticity as a function of heat treatment that would affect the material's residual stress state. We evaluate initial results obtained on two unidirectional (UD) systems: (1) continuous graphite (Gr) fiber reinforced Al, and (2) discontinuous SiC whisker reinforced Al. The requisite five elastic moduli C_{ij} for a UD system were obtained by measuring bulk acoustic velocities, first in the as-fabricated

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material. The engineering constants, e.g., longitudinal and transverse Young's moduli, were in good agreement with available tensile test data. The samples were then subjected to single cycle liquid nitrogen and elevated temperature excursions, and the elastic moduli remeasured at room temperature. Results indicate a significant effect on the residual stress state (specifically, a reduction in modulus) of Gr/Al, but no effect on SiC/Al.

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SUMMARY

The ability to make meaningful elastic moduli measurements ultrasonically is reported for potential application to the nondestructive evaluation of aluminum matrix composites. The materials were monitored for a change in elasticity as a function of heat treatment that would affect the material's residual stress state. We evaluate initial results obtained on two unidirectional (UD) systems: (1) continuous graphite (Gr) fiber reinforced Al, and (2) discontinuous SiC whisker reinforced Al. The requisite five elastic moduli C_{ij} for a UD system were obtained by measuring bulk acoustic velocities, first in the as-fabricated material. The engineering constants, e.g., longitudinal and transverse Young's moduli, were in good agreement with available tensile test data. The samples were then subjected to single cycle liquid nitrogen and elevated temperature excursions, and the elastic moduli remeasured at room temperature. Results indicate a significant effect on the residual stress state (specifically, a reduction in modulus) of Gr/Al, but no effect on SiC/Al.

J. R. Dixon

J. R. DIXON
By direction

PREFACE

The work reported here represents a portion of the research and development carried out by the Materials Division of the Research and Technology Department to evaluate the elastic properties of aluminum matrix composites. This work was supported by the NAVSEA Metal Matrix Composites Block Program SF 54 594594. This report summarizes ultrasonic test results and theoretical model calculations obtained on both continuous and discontinuous fiber reinforced aluminum composites in FY 78.

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ILLUSTRATIONS

TABLES

INTRODUCTION

The initial results of an ultrasonic study to nondestructively characterize metal matrix composite elastic properties are reported. Single cycle thermal treatment effects on the Young's and shear moduli, relevant to the end item fabrication process, are given for two specific unidirectional (UD) aluminum matrix composites. For one of these, SiC/Al, the as-cast billet moduli are compared to the extruded rod anisotropic values. Also, a comparison of the ultrasonically determined moduli with machine tensile values is made where possible.

The two UD metal composite systems addressed in this paper are (1) a continuous fiber Thornel 50 graphite reinforced 201 aluminum alloy (Th 50/201 Al); and (2) a discontinuous fiber silicon carbide whisker reinforced aluminum (SiC/Al). Both composites possess a nominal 30% by volume (30 V/o) reinforcement fiber array.

The motivation for thermal treatment studies is twofold. First, the as-fabricated residual stress state of a composite with constituent materials possessing different coefficients of thermal expansion is of concern. Secondly, the effects of net-shape material forming at elevated temperatures need to be evaluated.

Substantial efforts have been made in recent years to ultrasonically evaluate the engineering moduli of non-metal composites, with reasonable success.[1-6] However, The work in metal matrix composites has been much more limited.[7,8] However, the evolution of new inexpensive metal composites with very attractive features will accelerate the pace of metallic system development. Closer agreement can be expected between the ultrasonically determined dynamic moduli and the statically measured tensile values for metal rather than non-metal composites. This is due both to the strong viscoelastic frequency dependence of most non-metals and the effects of adiabatic heating (ultrasonic method) versus isothermal tests (static method). Both phenomena are negligible in non-viscous highly conductive metals. However, geometric dispersion (caused by an orderly array of fibers in the matrix) can dominate the apparent elasticity's frequency dependence in both metal and non-metal composites and, therefore, needs to be evaluated when making a direct comparison of dynamic and static test values.[9,10] Finally, we note that whether or not an absolute comparison can be confidently made between the ultrasonic and tensile values, the ultrasonic method is invaluable for accurately determining relative changes as a function of temperature, thermal treatment, and composition.

EXPERIMENTAL TECHNIQUE

Samples Used

Figure 1 illustrates the two composite systems investigated. The Gr/Al sample was available in single-ply plate form 2.6 mm thick.[a] The plate was formed from a multi-layer of precursor Gr/Al wires diffusion bonded with 0.15 mm of 2024 Al on both surfaces. The precursor wires, with a nominal diameter of 1 mm, were formed by liquid metal infiltration of graphite fiber bundles. The measured plate density was 2.44 gm/cm^3, indicating a net 30 V/o graphite fiber reinforcement. Parallel face samples approximately 3.0 mm in length were cut transverse to and at 45^o relative to the fiber direction. Figure 2 presents 30X and 100X magnifications normal to the transverse cut, illustrating the random fiber-aluminum matrix geometry.[b]

The SiC/Al sample was available in a 50 mm long by 40 mm diameter cast cylindrical billet with a 27 V/o reinforcement.[c] The SiC whisker dimensions are only approximately known: about 85% are particulate, and the remaining 15% have a nominal aspect ratio (ℓ/d) of 30, where d is $\sim$0.5 μm, after billet fabrication. The sample section taken from the 10:1 extruded bar was approximately 20 mm in length and 10 mm diameter. The extrusion process was expected to align the whiskers by shear viscous flow to produce a UD composite. From this extruded section, a 3.0 mm thick parallel face cut at 45^o to the extrusion direction was taken for three of the velocity measurements.

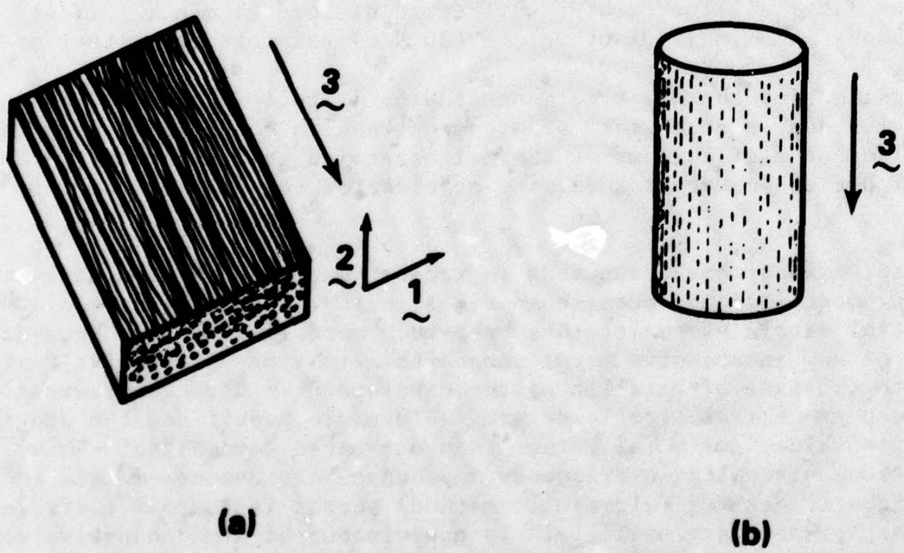

Figure 1. Pictorial Illustration of Unidirectional (a) Continuous Fiber Reinforced GR/Al Plate, and (b) Discontinuous Fiber (Whisker) Reinforced SiC/Al Extruded Rod. The Principal Fiber Direction is $\underset{\sim}{3}$, with the Plane of Isotropy Transverse to it.

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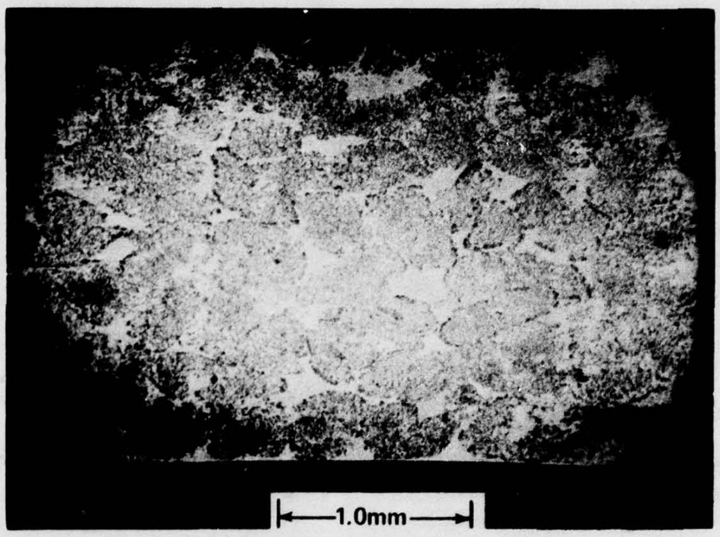
[a]Plate material fabricated by DWA Composite Specialities, Inc., Chatsworth, CA.
[b]Scanning electron micrograph taken by M. K. Norr, NSWC, Silver Spring, Maryland.
[c]Billet fabricated by A. P. Divecha, NSWC, Silver Spring, Maryland.

(a)



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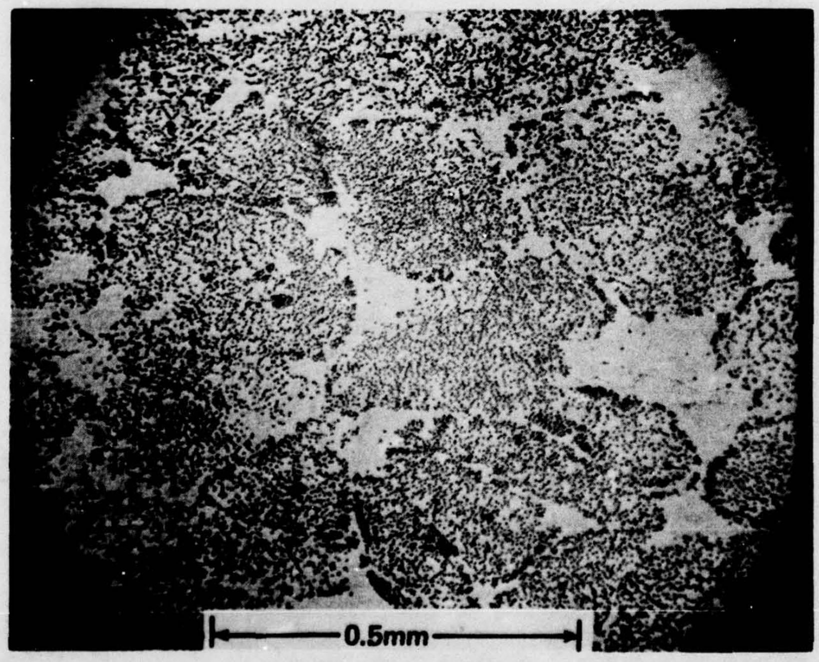

Figure 2. SEM Photographs of 30 V/o GR/Al Plate Sliced Transverse to the Fiber Direction (3) at Two Magnifications: (a) 30X Illustrating the Sample Plate Boundaries and Random Fiber Bundle Arrangements; and (b) 100X Illustrating the Al Infiltrated Array of Fibers in Each Bundle.

Ultrasonic System

Bulk acoustic velocity measurements of both longitudinal and shear waves were made in specific directions relative to the composite fibers. All measurements were made at room temperature by direct contact of transducer to sample or via a delay rod.

The experimental system used is shown in Figure 3. The sharp spike voltage excitation of a heavily damped PZT ceramic transducer for longitudinal waves, and of a $LiNbO_3$ single crystal for shear waves, produced broadband acoustic pulses with a nominal 10 MHz center frequency. Time of flight velocity measurements were made by pulse-echo overlap using the time delay of a 7000 series Tektronix oscilloscope. The digital readout time display provided a resolution of 1 ns for a precision $\sim$1:1000, with a like precision for the velocity. Minimum acoustic path lengths, defined by the sample dimensions, were 6 mm round trip measured to a resolution of 4 μm or $\sim$1:1000.

Data Analysis Approach

By means of the stress-strain constitutive relationships for orthotropic media, the elastic constants Cij were calculated knowing the material density ρ and the direction of acoustic propagation $\underset{\sim}{k}$ relative to the fiber direction $\underset{\sim}{3}$. Table I summarizes these relationships for a UD system, one that is transversely isotropic or specially orthotropic. There are five independent Cij as originally shown by Mason,[11] Musgrave[12] and others, but eight independent measurements are conveniently made for three directions of wave propagation relative to the fibers: $\underset{\sim}{k} \parallel \underset{\sim}{3}$, $\underset{\sim}{k} \perp \underset{\sim}{3}$, and $\measuredangle(\underset{\sim}{k}, \underset{\sim}{3}) = 45°$. The additional three measurements serve to check the other measurements.

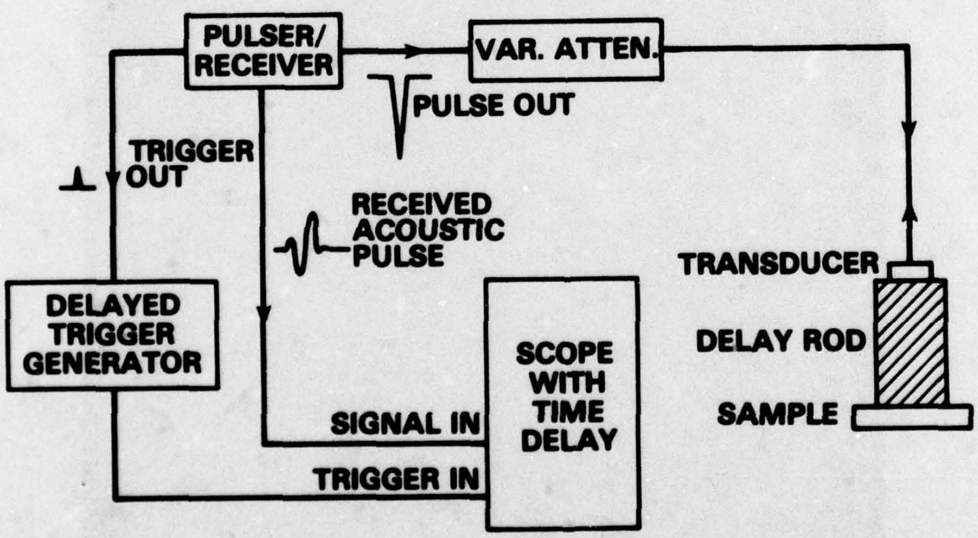

Figure 3. Broadband Ultrasonic System Used to Measure Sound Velocity By the Time-of-Flight Pulse-Echo Overlap Technique.

Table I. Equations Relating the Measured Ultrasonic Velocities Vn to the Elastic Moduli C_{ij} for a UD Composite with Fibers in the $\underset{\sim}{3}$ Direction

Equation #	Equation	Wave Type & Prop. Dir.[a]
1	$\rho V1^2 = C_{33}$	Long: $\underset{\sim}{k} \parallel \underset{\sim}{3}$
2	$\rho V2^2 = C_{11}$	Long: $\underset{\sim}{k} \perp \underset{\sim}{3}$
3	$\rho V3^2 = C_{44}$	Shear: $\underset{\sim}{k} \parallel \underset{\sim}{3}$, $\underset{\sim}{\epsilon} \perp \underset{\sim}{3}$
4	$\rho V4^2 = C_{44}$	Shear: $\underset{\sim}{k} \perp \underset{\sim}{3}$, $\underset{\sim}{\epsilon} \parallel \underset{\sim}{3}$
5	$\rho V5^2 = C_{66}$	Shear: $\underset{\sim}{k} \perp \underset{\sim}{3}$, $\underset{\sim}{\epsilon} \perp \underset{\sim}{3}$
6	$\rho V6^2 = 1/2 \left\{ \frac{C_{11} + C_{33} + 2C_{44}}{2} + \left[\left(\frac{C_{33} - C_{11}}{2} \right)^2 + \left(C_{13} + C_{44} \right)^2 \right]^{1/2} \right\}$	Quasi-Long: $\measuredangle(\underset{\sim}{k}, \underset{\sim}{3}) = 45^\circ$
7	$\rho V7^2 = 1/2 \left\{ \frac{C_{11} + C_{33} + 2C_{44}}{2} - \left[\left(\frac{C_{33} - C_{11}}{2} \right)^2 + \left(C_{13} + C_{44} \right)^2 \right]^{1/2} \right\}$	Quasi-Shear: $\measuredangle(\underset{\sim}{k}, \underset{\sim}{3}) = 45^\circ$, $\underset{\sim}{\epsilon}$ in plane $(\underset{\sim}{k}, \underset{\sim}{3})$
8	$\rho V8^2 = 1/2 \left(C_{44} + C_{66} \right)$	Shear: $\measuredangle(\underset{\sim}{k}, \underset{\sim}{3}) = 45^\circ$, $\underset{\sim}{\epsilon} \perp$ plane $(\underset{\sim}{k}, \underset{\sim}{3})$

[a] $\underset{\sim}{k}$ is the acoustic propagation vector, and $\underset{\sim}{\epsilon}$ is the shear wave polarization direction.

With the above-mentioned time resolution in the velocity measurement, the precision of the calculated Cii is ~4: 1000. Provided all controllable variables such as acoustic frequency and sample dimensions are kept constant, we can then monitor changes in Cii as a function of thermal treatment to better than 0.5%. The absolute uncertainty of this technique is estimated to be less than 1% for the Cii. For C_{13}, the number of variables involved in Equations 6 and 7 indicate a somewhat reduced precision. In fact, we have experienced the accuracy of C_{13} to be very much reduced, as have other investigators.[1,3]

Finally, to relate the composite bulk elastic properties so obtained to engineering quantities such as Young's modulus, we refer to Table II. The Emn, Gmn, and υ_{13} are, respectively, Young's moduli, shear moduli and Poisson's ratio, where the second subscript refers to the stress direction. The specific modulus type and property direction are given in the last column of Table II. Similar relationships exist for the other two Poisson's ratios pertinent to a UD composite. The precision of the longitudinal Young's modulus, E33, a principal quantity of interest for materials design, is estimated to be better than 2% with the above-described measurement technique.

Thermal Treatments Applied

Two sets of Gr/Al samples were subjected to separate thermal treatments. One set was subjected to a two-part cycle: first quenched in liquid nitrogen (LN_2), then heated at 260°C in a vacuum furnace for twenty minutes. This sample set was observed to have slightly expanded dimensions (<0.5%) after treatment. The second set was simply heated at 500°C under vacuum for twenty minutes. The LN_2 quenching step was applied to relieve the residual stress state of an as-fabricated plate material. The elevated temperature excursions simulate excursions the plate material experiences in net shape hardware fabrication and/or in service.

Only the 45° cut from the SiC/Al sample set was exposed to thermal treatment. It was heated at 500°C in the vacuum furnace for twenty minutes as was the Gr/Al sample set.

After thermal treatment, the elastic moduli of all specimens were remeasured ultrasonically at room temperature.

RESULTS

Gr/Al Composite

Table III provides a sample set of data for the 25 V/o Gr/Al specimens, together with the calculated Cij moduli. Additional C_{44} and C_{13} values (C_{44}-2, C_{13}-2, etc.) are calculated from the additionally measured velocities V6, V7 and V8. A computer program[d] has been written to calculate the Cij and the related engineering moduli for an arbitrary angle of the fibers relative to the acoustic propagation k. Table II above gives these relationships (Equations 6, 7, 8) for the special case of $\theta = 45°$.[2,3] In Table III, note that nine Cij values have been calculated from eight velocity measurements. The last value, C_{13}-3, was obtained by subtracting Equations 6 and 7 to solve for C_{13}.

[d]Program written by A. L. Bertram, NSWC, Silver Spring, Maryland.

Table II. Equations Relating Engineering Moduli to the Ultrasonically Determined Elastic Moduli C_{ij} for a UD Composite with Fibers in the $\underset{\sim}{3}$ Direction

Eq. #	Eng. Modulus	=	Function (C_{ij})	Type & Dir. Relative to Fibers
9	E_{33}	=	$C_{33} - \dfrac{C_{13}^{2}}{C_{11} - C_{66}}$	Young's: Long., Parallel to Fibers
10	E_{11}	=	$\dfrac{4\,C_{66}\left[C_{33}(C_{11} - C_{66}) - C_{13}^{2}\right]}{C_{11}\,C_{33} - C_{13}^{2}}$	Young's: Transverse to Fibers
11	G_{13}	=	C_{44}	Shear: In-Plane, Parallel to Fibers
12	G_{12}	=	C_{66}	Shear: In-Plane, Perpendicular to Fibers
13	ν_{13}	=	$\dfrac{C_{13}}{2\,(C_{11} - C_{66})}$	Poisson's Ratio: Stress Parallel to Fibers, Strain Measured in Transverse Plane

Table III. Sample Data Set for As-Fabricated 30 V/o Th 50/201 Al with a Density of 2.44 gm/cm^3

Velocity (mm/μs)	Equations Used from Table I	C_{ij} (GPa)[a]
V1 = 8.642	1	C_{33} = 182.
V2 = 4.040	2	C_{11} = 39.8
V3 = 2.783	3	C_{44}-1 = 18.9
V4 = 2.783	4	C_{44}-2 = 18.9
V5 = 2.073	3 & 8 or 4 & 8	C_{44}-3 = 18.8
V6 = 6.166	5	C_{66} = 10.5
V7 = 3.070	1, 2, 3, & 6	C_{13}-1 = 25.6
V8 = 2.450	1, 2, 3, & 7	C_{13}-2 = 25.5
	1, 2, 3, 6, & 7	C_{13}-3 = - 4.69

[a] 1 GPa = 10^9 N/m^2. To obtain Cij in units of 10^6 psi (or Msi), multiply # GPa by 0.145.

Some of the Cij calculations deserve special attention, in particular the C_{13} and C_{44} values. The large variation of C_{13}, even being negative, is not understood at this time but has been reported by other investigators.[1,3] Since the V7 measurement yielded the most meaningful and consistent results from sample to sample for both composite systems, C_{13}^{-2} was used throughout to calculate E_{11}, E_{33} and ν_{13} according to the relationships in Table II. Concerning the C_{44} values, although the particular data set in Table III shows excellent agreement for all three measurements, there was as much as a 9% difference observed in another sample between C_{44}^{-1} and C_{44}^{-2}. This is well in excess of the experimental precision of 0.5%. This may possibly be attributed to geometric dispersion since the acoustic wave length at 10 MHz is comparable to the average fiber spacing.[8,9] And these two shear measurements are made in one case with k parallel to the fibers, and the other with k perpendicular to the fibers.

Table IV summarizes the effects of two separate heat treatments on two sets of Gr/Al samples and compares the ultrasonic (UT) values with theoretical model predictions for as-fabricated plate before heat treatment. The UT as-fabricated values were calculated according to Equations 9-13 using the raw data of Table III. Model predictions assumed perfect bonding of the graphite fibers to the Al matrix,[13,14] in addition to static stress conditions. (See Appendix for the relevant formulae of the variational bounds model by Z. Hashin.) In spite of this, the second and third columns of Table IV show very good agreement between experiment and theory, except for Poisson's ratio ν_{13}. At this point, we only note that ν_{13} is proportional to C_{13}, a quantity which itself presents interpretation difficulties. Finally, we note that E_{33} modulus was in good agreement with the range of values (131 to 165 GPa) obtained on a series of tensile specimens taken from the same sample plate. In addition, the ultrasonic E_{11} and G_{13} values were in good agreement with available static test results obtained on similar plate material.[e]

A comparison of the last two columns of Table IV with the first column shows a measureable reduction in Young's moduli, in particular E_{33}, after thermal treatment. The changes in the shear moduli, however, are negligible. Lastly, although the experimental precision is sufficient to detect a real change in Poisson's ratio, the strong dependence on C_{13} (Equation 13) clouds any significance that might be attached to its apparent increase.

Possible physical mechanisms for the reduction in E_{33} can only be conjectured at the present time. The possible formation of Al_4C_3 at the Gr/Al interface is unreasonable for such short duration heat treatments (twenty minutes) at 260°C and 500°C.[15-17] It is more plausible to consider the effects on the residual stress state of the material, although reductions on the order of the 10% measured are unexpected.[18] Future heat treatments will help to elucidate the mechanism.

SiC/Al Composite

Table V compares ultrasonic moduli of 27 V/o SiC/Al before and after extrusion, including theoretical model predictions for the (presumed) unidirectional extruded rod. The reinforced cast billet values in the first column represent a 70% increase in stiffness over the isotropic Al matrix values of 70 and 27 GPa, respectively, for E and G. The last two columns show good agreement between the ultrasonic values

[e]Tensile data provided by NETCO, Long Beach, California.

Table IV. Ultrasonic Moduli (in GPa) of 30 V/o Th 50/201 Al Before and After Two Separate Heat Treatments, Including Theoretical Model Predictions

Eng. Modulus	As-Fabricated Plate		Post Heat Treatment	
	UT	Model[a]	500°C	LN2 Quench & 260°C[b]
E_{33}	160.	166.	148.	141. ± 3.
E_{11}	29.7	32.6	29.7	26.9
G_{13}	18.6	18.5	18.6	19.3
G_{12}	10.3	10.8	11.0	9.6
ν_{13}	0.43	0.34	0.47	0.53

[a] Continuous fiber reinforcement model by Hashin.[14]

[b] Relatively large error in precision due to different specimen used.

and theoretical model predictions. The difficulties experienced in the calculation for C_{13} for Gr/Al were not encountered with the SiC/Al data. The discontinuous fiber reinforcement model assumes 100% alignment of the elongated whiskers in the direction of extrusion.[19,20] In addition, equal longitudinal and transverse Young's moduli for whisker elasticity, estimated at 480 GPa, is assumed. More recent evidence that the value may be closer to 700 GPa might be cause for ultrasonically measuring a modulus value E_{33} larger even than that predicted by a model assuming perfect whisker-matrix bonding.

The high percentage of particulate SiC with $\ell/d \sim 1$ is reflected by the high elastic isotropy (nearly equal shear moduli G_{12} and G_{13}) measured. The available tensile data on this sample, a three-test average value of 117 GPa for E_{33}, showed reasonable agreement with the ultrasonically determined value of 132 GPa.[f]

Table VI presents data on SiC/Al before and after a twenty-minute heat treatment at 500°C. The calculated engineering moduli are bracketed to indicate that the before-heat treatment data was necessarily used in part for these calculations. Since only the 45° cut (to the extrusion direction) sample was available for heat treatment, only the V6, V7 and V8 measurements could be made on the heat treated specimen. However, the consistency of the three measurements, indicating a constant or slightly increasing stiffness, lends confidence to the bracketed values. In addition, data on several other heat treated specimens of extruded SiC/Al has consistently demonstrated a slight increase in modulus value with elevated temperature excursions. Both a more comparable thermal coefficient of expansion between SiC and Al, and the short discontinuous fiber nature of SiC reinforcement, corroborates the lesser heat treatment effects observed on the elasticity of SiC/Al than on Gr/Al.

CONCLUSIONS AND PLANS

The most significant result is the effect of thermal treatment on the two metal matrix composites' elasticity: a measurable reduction in stiffness for Gr/Al but no reduction for SiC/Al. Secondly, the good agreement of the ultrasonic values with machine tensile test data and model predictions demonstrates that the ultrasonic technique can be used to meaningfully characterize the material moduli.

An unresolved difficulty is the inconsistency of the C_{13} calculations from the 45° velocity measurements for Gr/Al. This principally reflects itself in questionable values for the Poisson's ratio which are anomalously large in some cases. The small increase upon extrusion in the longitudinal modulus E_{33} for SiC/Al indicates that the use of better quality reinforcement whiskers should improve that material's already attractive properties.

Future investigations will be first to examine the effects of repeated heat treatments on the material's elasticity. Secondly, multi-ply Gr/Al laminates stacked at specific fiber angles (to improve the transverse properties) will be studied. Finally, some long-range goals for the program include making a reliable nondestructive estimate of Young's modulus for a UD composite by means of a simple longitudinal velocity measurement, and to relate the material's strength to ultrasonic parameters.

[f] Tensile data provided by C. R. Crowe, NSWC, Silver Spring, Maryland.

Table V. Ultrasonic Moduli (in GPa) of 27 V/o SiC/Al Before and After Extrusion, Including Theoretical Model Predictions

Eng. Modulus	CAST BILLET	EXTRUDED ROD	
	UT	UT	Theor. Model[a]
E_{33}	116.	132.	124.
E_{11}	-	116.	117.
G_{13}	44.8	44.8	45.5
G_{12}	-	43.7	40.7
ν_{13}	0.29	0.29	0.30

[a]Discontinuous fiber reinforcement model by Halpin and Tsai.[19]

Table VI. Ultrasonic Moduli (in GPa) of 27 V/o SiC/Al Extruded Rod Before and After Heat Treatment at 500°C for 20 minutes

Eng. Modulus	BEFORE	AFTER
E_{33}	132.	[134.][a]
E_{11}	116.	[117.]
G_{13}	44.8	[44.8]
G_{12}	43.7	[43.7]
ν_{13}	0.29	[0.28]

[a] Brackets [ ] indicate values calculated using BEFORE heat treatment data in part.

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Appendix

Following are the formulae used to obtain the theoretical model predictions by Hashin reported in Table IV for as-fabricated Gr/Al plate. The effective elastic moduli for a unidirectional continuous fiber composite consisting of transversely isotropic fibers and matrix are indicated by stars: E_A^*, E_T^*, etc.

The particular relationships peculiar to a Gr/Al composite are E_T^* and G_T^* given below in equations (A-2) and (A-4). These relationships represent the lower bound solutions of the variational bounds method based on the composite cylinder assemblage model.[14] These results incorporate the boundary condition presented by a matrix which has greater transverse moduli than the fiber.

In all the equations, on the right hand side the suffix 1 denotes matrix and 2 denotes fibers; and A denotes the axial or 3 direction, and T denotes the transverse direction. The bulk modulus is k, and the respective volume fractions of matrix and fiber are v_1 and v_2. Equations (A-2) and (A-4) are for the case that $G_{T1} > G_{T2}$ and $E_{T1} > E_{T2}$.

$$E_{33} \equiv E_A^* = E_{A1}v_1 + E_{A2}v_2 + \frac{4(\nu_{A2} - \nu_{A1})^2 v_1 v_2}{v_1/k_2 + v_2/k_1 + 1/G_{T1}} \tag{A-1}$$

$$E_{11} \equiv E_T^* = \frac{4k^*G_T^*}{k^* + m^*G_T} \tag{A-2}$$

$$G_{13} \equiv G_A^* = G_{A1}\,\frac{G_{A1}v_1 + G_{A2}(1 + v_2)}{G_{A2}(1 + v_2) + G_{A2}v_1} = G_{A1} + \frac{v_2}{\dfrac{1}{G_{A2} - G_{A1}} + \dfrac{v_1}{2G_{A1}}} \tag{A-3}$$

and

$$G_{12} \equiv G_T^* = G_{T1}\left[1 + \frac{(1 + \beta_1)v_2}{\rho - v_2 \quad 1 + \dfrac{3\beta_1^2 v_1^2}{\alpha v_2^3 - \beta_1}}\right] \tag{A-4}$$

$$\nu_{13} \equiv \nu_A^* = \nu_{A1}v_1 + \nu_{A2}v_2 + \frac{(\nu_{A2} - \nu_{A1})(1/k_1 - 1/k_2)v_1v_2}{v_1/k_2 + v_2/k_1 + 1/G_{T1}} \tag{A-5}$$

where:

$$k_2 = \frac{E_{A2}\ E_{T2}}{2\ E_{A2}\ (1 - \nu_{T2}) - 4\ E_{T2}\ \nu_{A2}^2} \tag{A-6}$$

$$k^* = k_1 + \frac{v_2}{\dfrac{1}{k_2 - k_1} + \dfrac{v_1}{k_1 + G_{T1}}} \tag{A-7}$$

$$m^* = 1 + \frac{4k^*\nu_A^{*2}}{E_A^*} \tag{A-8}$$

$$\alpha = \frac{\beta_1 - \gamma\beta_2}{1 + \gamma\beta_2} \tag{A-9}$$

$$\rho = \frac{\gamma + \beta_1}{\gamma - 1} \tag{A-10}$$

$$\gamma = \frac{G_{T2}}{G_{T1}} \tag{A-11}$$

$$\beta_1 = \frac{k_1}{k_1 + 2G_{T1}} \tag{A-12}$$

$$\beta_2 = \frac{k_2}{k_2 + 2G_{T2}} \tag{A-13}$$

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